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LA-UR -90-2801

SEP 0 7 1990

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LA-UR--90-2801

DE90 016489

TITLE MECHANICAL AND MICROSTRUCTURAL RESPONSE OF Ni_3Al AT HIGH STRAIN RATE AND ELEVATED TEMPERATURES

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SUBMITTED TO International Conference on Shock-Wave and High-Strain-Rate Phenomena in Materials, EXPLOMET 1990, San Diego, CA, August 12-17, 1990.

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In this paper, the effect of strain rate and temperature on the substructure evolution and mechanical response of Ni_3Al will be presented. The strain rate response of Ni_3Al was studied at strain rates from 10^{-3} s^{-1} (quasi-static) to 10^4 s^{-1} using a Split Hopkinson Pressure Bar. The Hopkinson Bar tests were conducted at temperatures ranging from 77K to 1273K. At high strain rates the flow strength increased significantly with increasing temperature, similar to the behavior observed at quasi-static rates. The work hardening rates increased with strain rate and varied with temperature. The work hardening rates, appeared to be significantly higher than those found for Ni270. The substructure evolution was characterized utilizing TEM. The defect generation and rate sensitivity of Ni_3AL are also discussed as a function of strain rate and temperature.

I. INTRODUCTION

While the influence of strain rate on the structure/property response of a variety of metals and alloys has been extensively studied, the effect of strain rate on ordered alloys remains largely unknown. Extrapolating trends in mechanical properties which are well known in conventional alloys to ordered systems is complicated by the unusual temperature dependence of the flow strength in some intermetallics such as Ni_3Al. In most materials the flow stress decreases with increases in temperature over all strain rate regimes. In Ni_3Al and other ordered intermetallics, the anomalous yield strength behavior has been recognized for many years [1-6]. The strain rates at which previous investigations have been conducted on Ni_3Al have been limited to rates less than 0.1 s^{-1} [1-6]. In these strain rate regimes the flow stress increased with temperature to about 1000K and then decreased with further increases in temperature [3-6]. The cause of the anamolous flow stress behavior in Ni_3Al is currently the topic of wide discussion. Most agree that this behavior is due to a change in planarity of the mobile dislocations and the locking of these dislocations by the kinks that form during cross-slip.

At high strain rates, $1 < \dot{\epsilon} < 10^4$ s^{-1}, the mechanism controlling deformation in FCC metals is thought to be thermally activated glide [7-10]. Although Ni_3Al is an ordered FCC metal, the anamolous flow stress behavior should still be observed at high strain rates if the rate controlling mechanism is thermally activated. The impetus behind this research was to examine whether or not this anamolous behavior was observed at high strain rates.

II. EXPERIMENTAL

Ni_3Al obtained from Idaho National Engineering Laboratory, with a composition of 75.9 at.% Ni, 24.1 at.% Al and 0.095 at% B, was used in this study [11]. This alloy is a powder metallurgy product fabricated by Homogeneous Metals of Clayville, NY. The powders were produced by vacuum gas atomization, sealed in evacuated steel cans and extruded at 1100 °C. Compression samples were cut with a wire EDM. The samples were annealed at 1100°C in sealed quartz tubes or in an inert atmosphere and were quenched in water. This heat treatment yielded an equiaxed microstructure with a 40 μm grain size. Tests were also conducted on samples that were furnace cooled, but no discernable differences in the mechanical behavior were found. Quasi-static compression tests were conducted on a screw-driven load frame at strain rates of 0.001 and 0.1 s^{-1}. Dynamic tests, strain rates above 10^3 s^{-1}, were conducted in a split-Hopkinson pressure bar [12]. High temperature tests were performed in a vacuum furnace mounted on the split-Hopkinson bar [13]. TEM foils were cut from the samples after testing using a low speed diamond saw. The foils were thinned mechanically with 600 grit paper and then jet polished in a Fischione dual jetpolisher. The polishing conditions were -40°C, 60 volts and 10 mA in an ethanol-4% perchloric acid solution. The foils were examined in either a Phillips CM30 AEM or a JEOL 2000EX.

III. RESULTS AND DISCUSSION

At a strain rate of 3000 s^{-1}, the flow stress at six percent strain increases with temperature, between 300K and 1083K. (Fig. 1). The flow stress at 77K is slightly higher than that found at room temperature. This behavior is similar to that found in the low strain rate regimes [2-6]. Above 1083K the flow stress at six percent strain saturates and only decreases slightly at higher strains. This saturation is significantly different than that observed at this temperature in the quasi-static strain rate regimes for Ni_3Al, but similar to that seen in pure metals at high strain rates. At larger strains, in the high temperature range, the flow stress reaches saturation, i.e. Ni_3Al shows no work hardening, and the flow stress falls to levels less than those found at the same strains at lower temperatures.

The strain rate sensitivity of Ni_3Al varies considerably with temperature. At 77K, Ni_3Al shows no strain rate sensitivity over strain rates ranging from 2000 to 7000 s^{-1}. Both at room temperature and 545°C, Ni_3Al shows significant increases in flow stress with strain rates ranging from 2000 s^{-1} to 8000 s^{-1}. The strain rate sensitivity at room temperature, however, is lower than that found at 545°C over an equivalent increase in strain rate. This increase in strain rate sensitivity with temperature is commonly observed in other metals [14].

The variations in work hardening rate with strain rate and temperature, perhaps because of the anamolous flow behavior, are in some respects different than those seen in Ni. For strain rates greater than 2000 s^{-1} at 77K the strain hardening rate ($d\sigma/d\epsilon$) increases with increasing flow stress (Fig. 2a). This increase in strain hardening rate is also observed in the lower strain rates at room temperature, although the rate of increase with flow stress is not as high as that seen at 77K (Fig. 2a). At temperatures between room temperature and 473K and a strain rate of 3000 s^{-1}, the strain hardening rate remains relatively constant with increasing flow stress, but decreases slightly with temperature (Fig. 2b). At temperatures greater that 823K the strain hardening rate decreases with increasing flow stress, which is similar to the behavior of most metals [7-8].

At room temperature the strain hardening rate increases with increasing strain rate but is insensitive to flow stress (Fig. 3). The value of the strain hardening rate for Ni_3Al at room temperature and quasi-static strain rates, is approximately 10% higher than that found for Ni270 [9]. The strain hardening rate for high strain rates is found to be considerably higher by about 40%, ($\mu/100$ @ $\epsilon = 3000\ s^{-1}$ for Ni_3Al versus $\mu/145$ for Ni270) than that found for Ni270 [9-10]. Some caution must be taken in making a direct comparisons of work hardening rate values between Ni_3Al and Ni because of possible differences in internal structure (dislocation density, grain size, etc.), the dependence of strain hardening on strain rate [6-9] and the observation that ordered materials strain harden at a higher rate than non-ordered materials [4-6]. At 545°C the strain hardening rate also increases with strain rate. At strain rates less than 5000 s^{-1} the strain hardening rate decreases with increasing flow stress and falls in the region of decreasing work hardening rates shown in Figure 2c. At strain rates higher than 5000 s^{-1} the strain hardening rate remains relatively constant with increasing flow stress.

A. TEM Observations

After annealing the Ni_3Al was relatively dislocation free. Samples deformed at a strain rate of 0.001 s^{-1} had an even distribution of dislocations which were predominately of the $<110>\{111\}$ type (Fig. 4a). Stacking faults were also present in the material deformed at this strain rate. The sample deformed at 3000 s^{-1} showed a similar dislocation structure but a higher dislocation and stacking fault density. At a strain rate of 8000 s^{-1} the dislocation density was quite high and coarse planar slip bands lying on $\{111\}$ planes were observed in many grains (Fig. 4b).

The variation in dislocation structure with temperature at a strain rate of

3000 s^{-1} is similar to that found at quasi-static rates. At liquid nitrogen temperature the deformation is limited to <110>{111} type dislocations with a very high density of stacking faults. This high density of stacking faults most likely contributes to the lack of strain rate sensitivity. At 673K, <110>{111} type dislocations are predominate but <110>{100} type dislocations and stacking faults are also observed. The stacking fault density decreases with increasing temperature, with only a few stacking faults observed in the material deformed at 1273K. In the samples deformed at 973K and 1273K the number of <110>{001} dislocations increases, however the predominate dislocations are still <110>{111} type (Fig. 4d). Also observed in these samples were <110>{110} dislocations; a configuration reported by Caron et.al. [15]. The materials deformed at higher temperatures also have considerably lower dislocation densities. Since the material is recovering faster than the dislocations can accumulate, it is difficult to say which dislocations are controlling the deformation behavior at the high strain rates and high temperatures. One might speculate that if the <110>{001} dislocations were controlling deformation then recovery in this system is much more rapid than in the <110>{111} system. A study is currently under way to examine this possibility.

IV. SUMMARY

The anomolous flow strength reported for Ni_3Al at quasi-static strain rates was also observed at strain rates of 3000 s^{-1}. The work hardening rate found in Ni_3Al was higher than Ni270 at room temperature. The temperature dependence of the work hardening rate versus flow stress was qualitatively similar to that found in most metals; with temperature increases the work hardening rate decreases. Ni_3Al had a strain rate sensitivity that was higher than that observed in Ni270.

Acknowledgements - The authors would like to acknowledge the assistance of W. Wright and C. Trujillo for performing the Hopkinson bar tests and and M. Lopez for conducting the quasi-static tests. This research was performed under the auspices of the United States Department of Energy.

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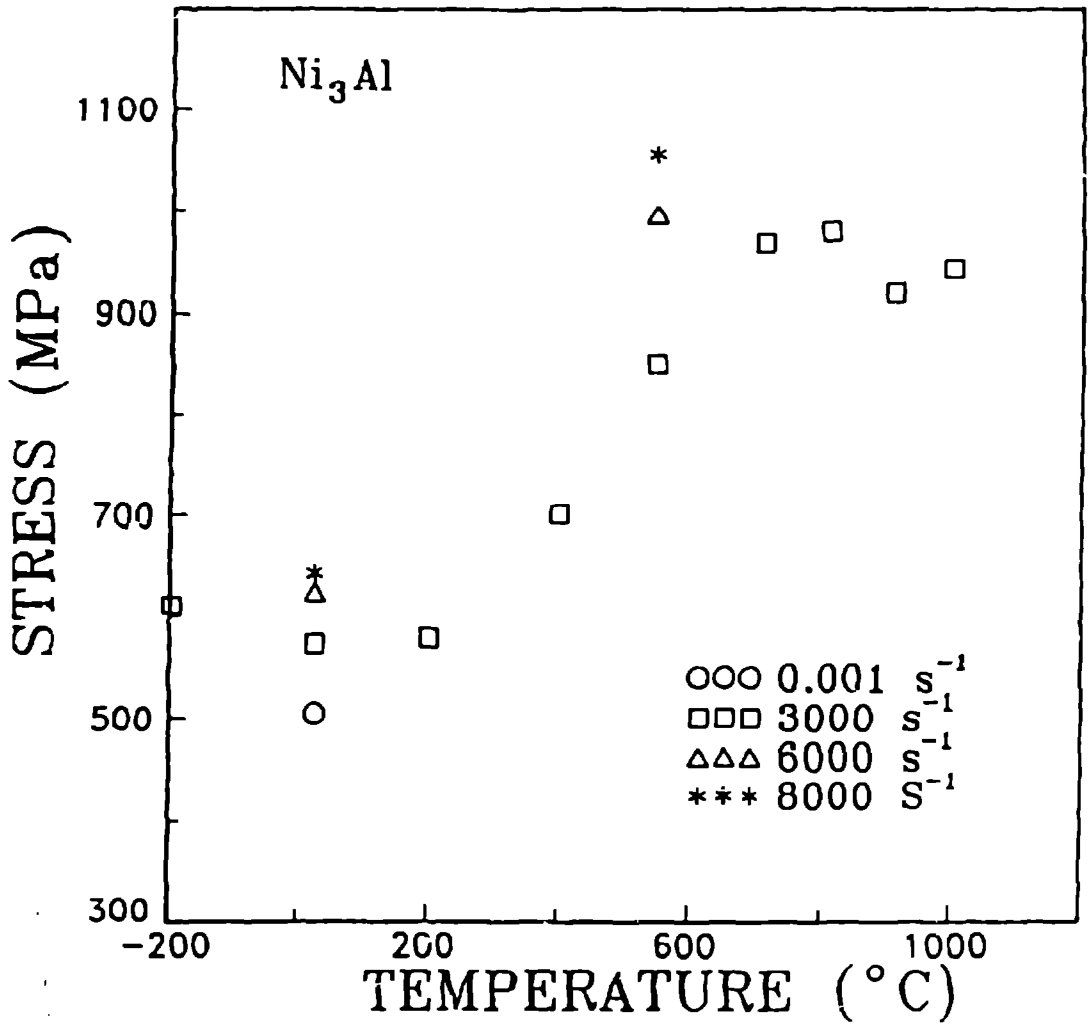
FIG. 1 *Flow stress at 6% strain versus temperature for Ni_3Al with 0.095at% B.*

FIG. 2 *Work hardening rate versus flow stress for Ni_3Al with 0.095 at% B deformed at a) 77K $\epsilon > 2000 s^{-1}$ and 300K low ϵ b) 300K, 473K c) 677K, 823K, 983K.*

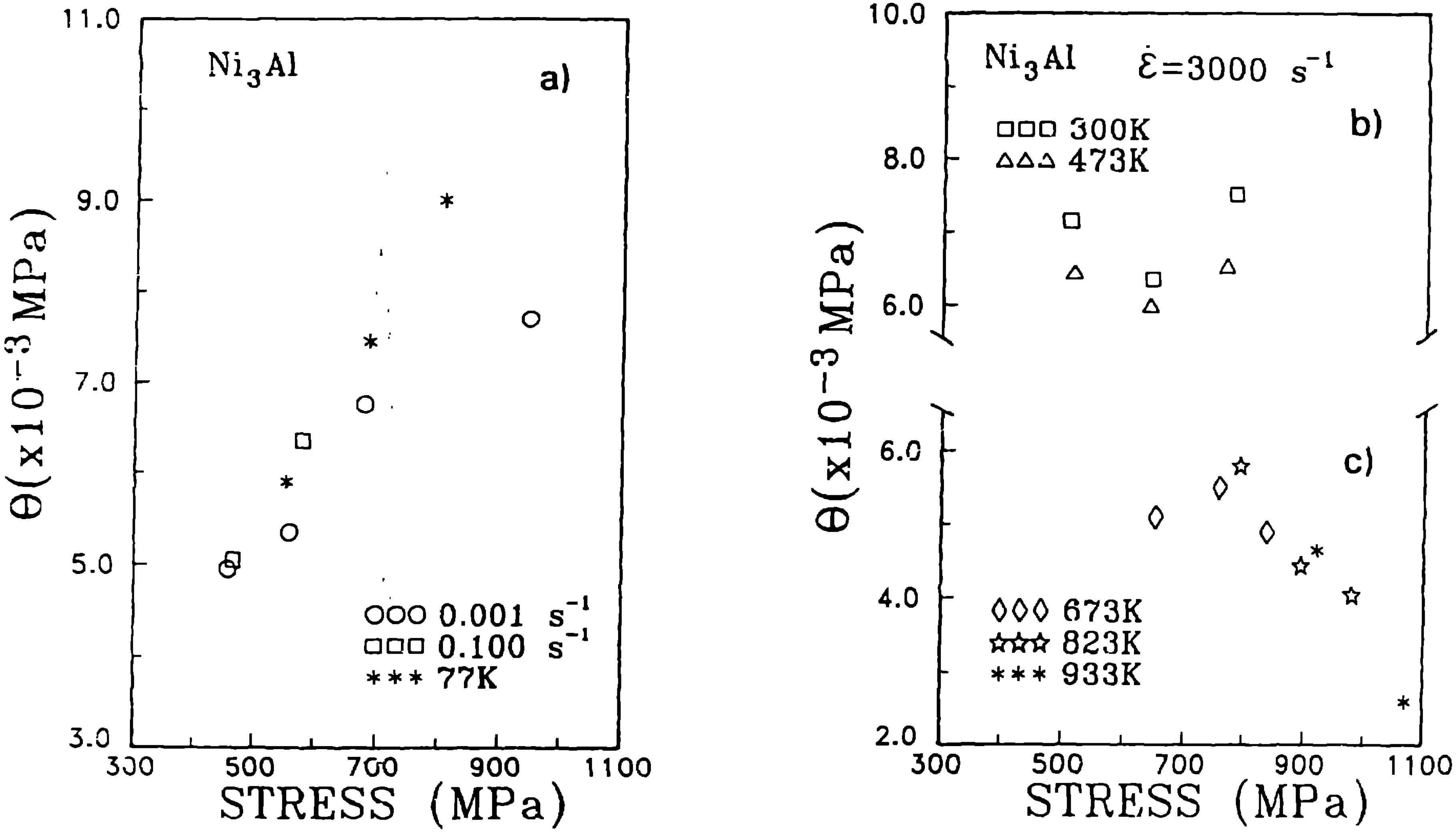
FIG. 3 *Work hardening rate versus flow stress for Ni_3Al deformed at various strain rates for a) 300K b) 823K.*

FIG. 4 *TEM brightfield of Ni_3Al deformed at a) 300K and $\epsilon = 0.001 s^{-1}$ b) 300K, $\epsilon = 8000 s^{-1}$ c) (Dark field) 77K, $\epsilon = 3000 s^{-1}$ d) 1273K, $\epsilon = 3000 s^{-1}$.*



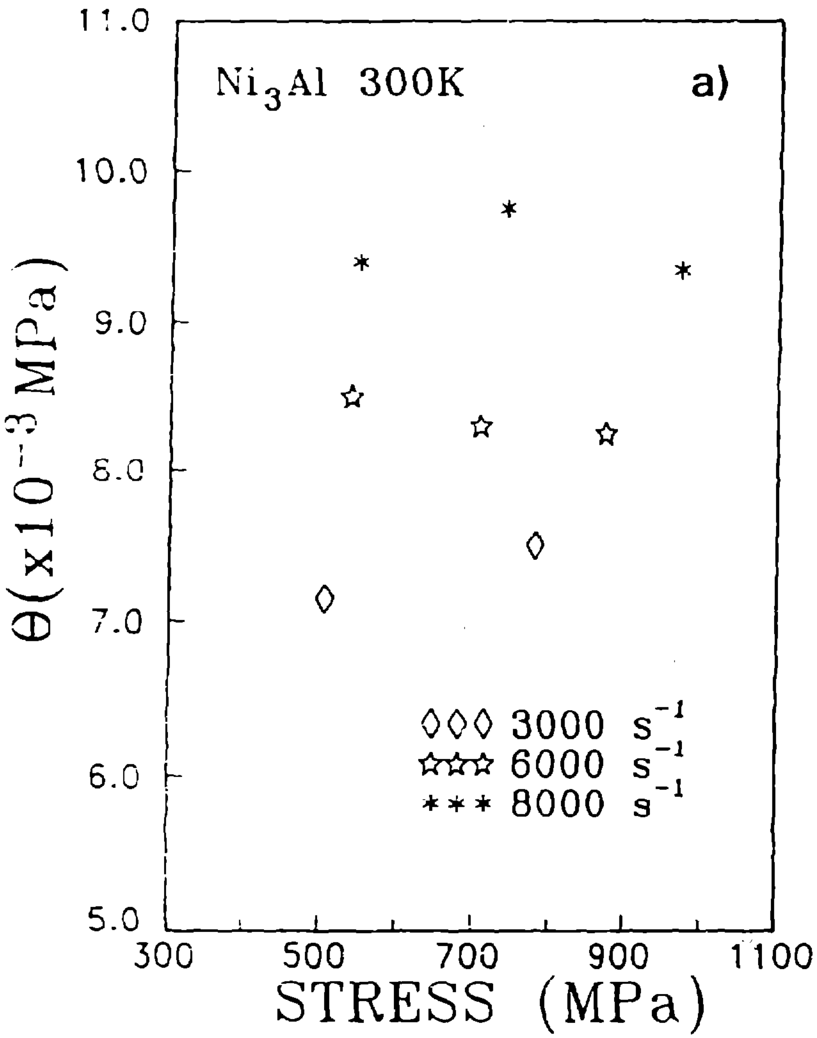
Ni_3Al
STRESS (MPa)
1100
900
700
500
300
-200
200
600
1000
TEMPERATURE (°C)
○○○ 0.001 s^{-1}
□□□ 3000 s^{-1}
△△△ 6000 s^{-1}
*** 8000 S^{-1}

FIG 1



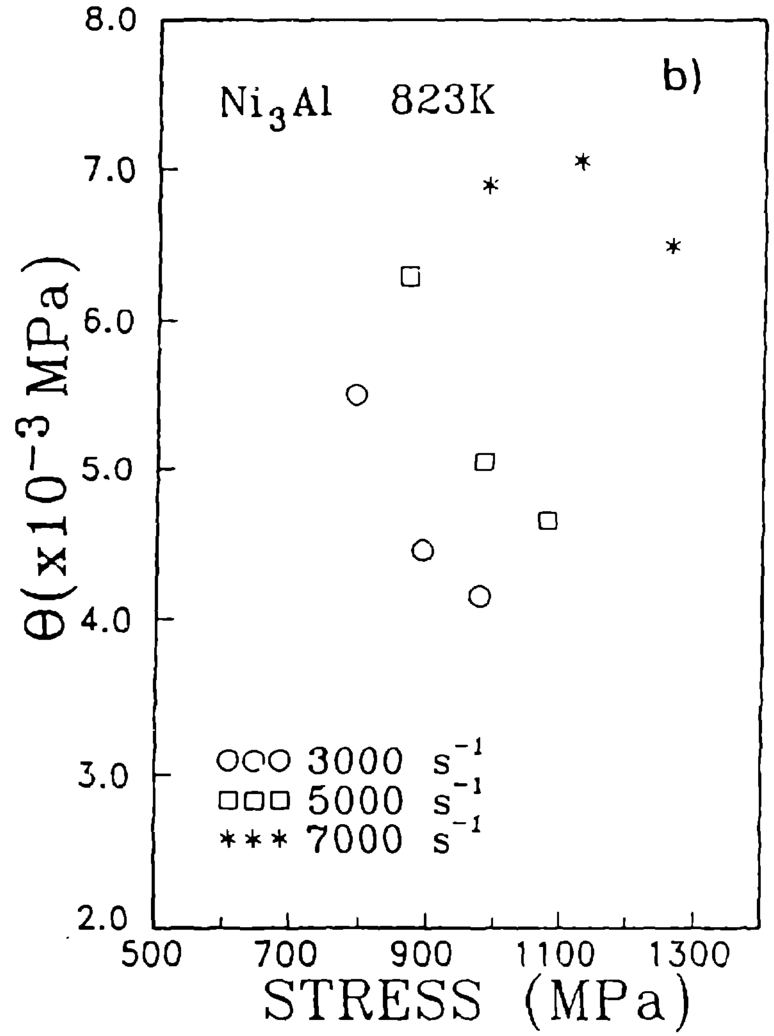
Ni3Al
a)
11.0
9.0
7.0
5.0
3.0
300
500
700
900
1100
STRESS (MPa)
Θ(x10⁻³ MPa)
0.001 s⁻¹
0.100 s⁻¹
77K
Ni3Al
ε̇=3000 s⁻¹
b)
300K
473K
10.0
8.0
6.0
c)
6.0
4.0
2.0
673K
823K
933K

Fig 3a



Ni3Al 300K
a)
◊◊◊ 3000 s^-1
☆☆☆ 6000 s^-1
*** 8000 s^-1
Θ(x10^-3 MPa)
STRESS (MPa)

Fig 3b



Ni3Al 823K
b)
○○○ 3000 s^-1
□□□ 5000 s^-1
*** 7000 s^-1
Θ(x10^-3 MPa)
STRESS (MPa)

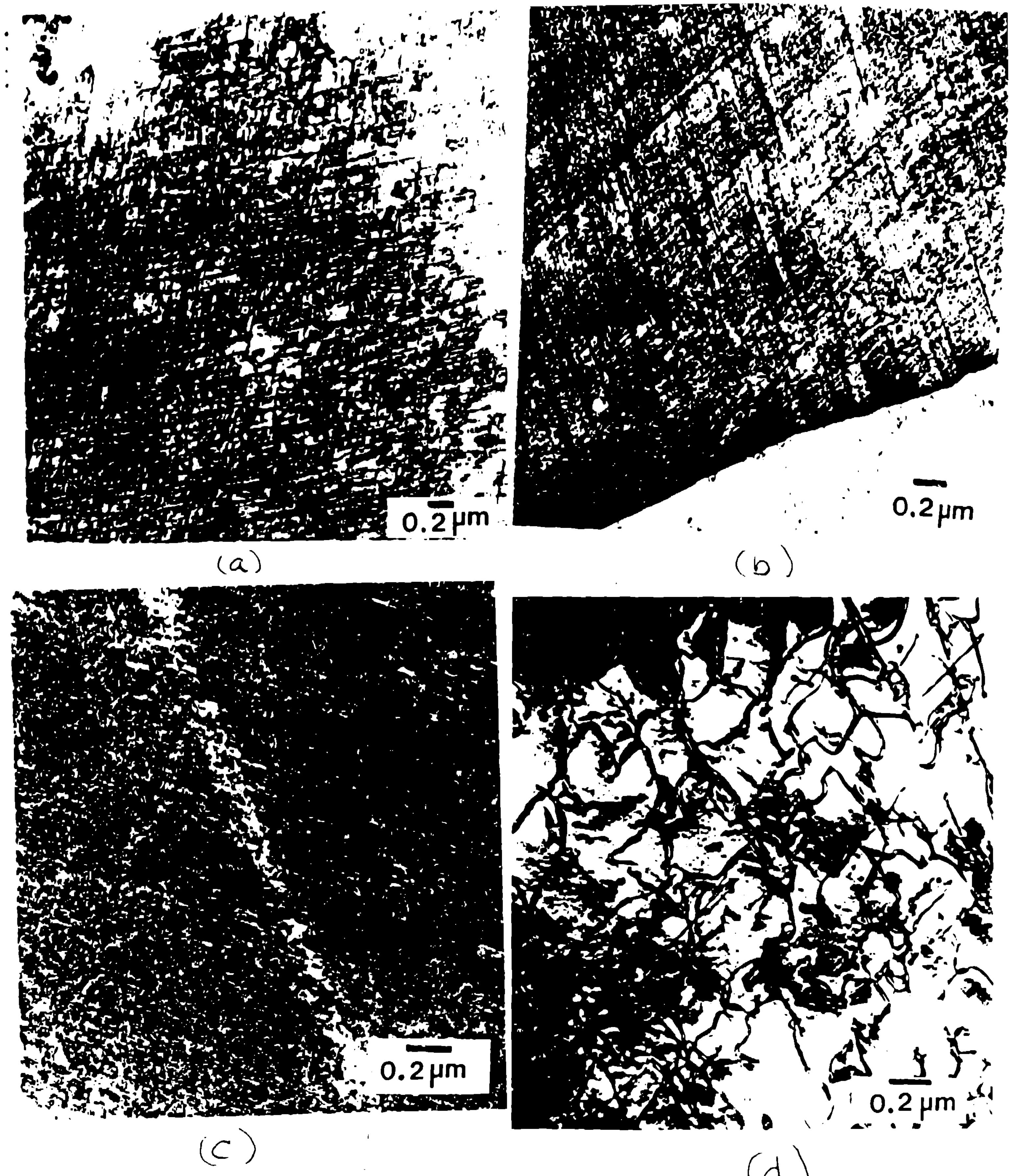


FIG 4.